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INDENTATION: DEFORMATION AND FRACTURE PROCESSES

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A summary of recent developments in the study of indentation processes in glass is presented. Attention is focussed on ideally "sharp" indenters in which the contact deformation contains both reversible and irreversible components. The relative amounts of these two components are determined by the ratio of hardness to elastic modulus, and are directly measurable from the depth recovery of the impression. At high loading rates the plastic work rate may be sufficient to cause local surface "melting".

Above some threshold in the loading, cracks initiate from the deformation zone, where strong stress concentrations exist. This threshold is found to be rate dependent, with a tendency to an increasingly delayed, post-indentation pop-in at lower peak loads. The presence of water strongly diminishes the delay times. Once initiated, the cracks grow spontaneously into their well-defined radial/median and lateral configurations, to a size determined primarily by the material toughness. Again, these cracks continue to extend with time in water-containing environments. The role of residual stresses in driving these fracture processes is a vital element of the mechanics description.

Microscopic examination of the indentation patterns provides useful information on the fundamental micromechanisms of deformation and fracture in glasses. The deformation involves a "shear fault" mode, at least in glasses with a significant network modifier content. These faults act as the precursors to crack initiation. The cracks show growth characteristics, e.g. a tendency to reversibility, indicative of a classical bond-by-bond rupture process at their tips.

I. INTRODUCTION

Indentation is now widely adopted as a tool for characterizing the intrinsic mechanical properties of glasses and other brittle materials.[1–6] The contact of a hard indenter with a specimen surface generates a stress field which, although complex, gives rise to well-defined deformation-fracture patterns. With the proper choices of contact test conditions a wide range of mechanical responses can be studied: the issues of elastic vs inelastic (plasticity, viscosity or densification) modes in the deformation, and of crack initiation vs propagation in the fracture, are pertinent examples. It is this flexibility, coupled with the more obvious attributes of simplicity and control in the experimental procedure, which appeals to those who seek to understand the micromechanics of the various possible deformation and fracture processes at the fundamental level.

In this paper we shall survey some of the more recent advances in the indentation analysis of glasses. We shall direct most of our attention to one particular contact configuration, that of a fixed-profile "sharp" indenter loaded axially on to the specimen. This is not to suggest that other well-known contact configurations are unworthy of consideration: the damage patterns produced using spherical ("blunt") indenters, and in contacts with a tangential component in the loading, have certainly aroused a great deal of interest in the past.[1] However, our concern here will be with material properties rather than with details in the contact geometry, and it is the simple pyramidal hardness indenters (Vickers, Knoop) which appear to be the most versatile in this regard. Sharp-indenter damage also provides us with greater insight into the physical characteristics of real flaws in glass, although, again, this aspect will not be explicitly addressed in the present work. Some of these other topics are discussed in other papers in this volume.

Our presentation will be in two main parts: in the first, we ask when and where in the contact event the processes of deformation and fracture occur (mechanics); in the second, we ask how and why (mechanisms).

II. INDENTATION MECHANICS IN SHARP CONTACT

The essential features of the classical deformation-fracture pattern produced by sharp indenters are shown in Fig. 1.[7–13] The severity of the contact is most conveniently characterized by the normal loading force P and/or the penetration z. A "plastic" enclave, embedded in an elastic far field, develops about the sharp point (and edges) of the indenter as it enters the underlying material. Dimensions a and b, representing the scale of the hardness impression and plastic zone, respectively, are used to quantify the extent of the deformation. Two distinctive crack types, median/radial

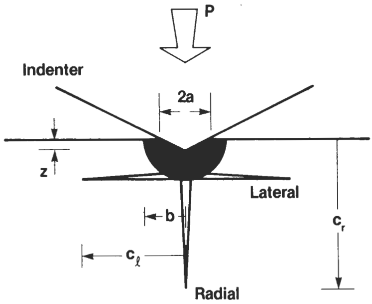

Fig. 1. Schematic of deformation-fracture pattern associated with a sharp indenter.

(hereafter referred to simply as "radial") and lateral, generate from the deformation zone: the former, characterized by dimension c_r, lie on planes containing the load axis and an impression diagonal; the latter, characterized by c_ℓ, lie on subsurface planes closely parallel to the specimen surface. Sometimes cone cracks (not shown in Fig. 1) are also generated.

The objective of any general mechanical description of the contact problem is to relate the quantities defined in Fig. 1 to appropriate material parameters, such as elastic (Young's) modulus E, hardness H, and toughness K_c.

A. DEFORMATION ZONE

We have indicated that the deformation beneath a sharp indenter contains both a reversible and an irreversible component. A detailed stress analysis of contact configurations of this kind is a formidable task. The simplest models are based on the concept of a pressurized internal cavity,[14] with immediate plastic and remote elastic surround volumes simulating the enclave geometry of Fig. 1. Such models, despite serious shortcomings,[15] are valuable for the way they bring out the interdependence of the two deformation components. Accordingly, the ratio of hardness to elastic modulus, H/E, emerges as a key parameter in the description; the greater this ratio, the greater the role of the elastic component in the contact properties. Glass, with a value of H/E approaching 0.1, lies at the top end of the materials spectrum in this regard, a fact attributable to the relatively high rigidity of the silicate tetrahedral network.[16]

An interesting manifestation of this strong tendency to an elastic response in the contact deformation of glass is a correspondingly strong depth recovery at the unloaded impression.[17,18] This is most conveniently represented on a plot of the load-penetration function P(z) over a full

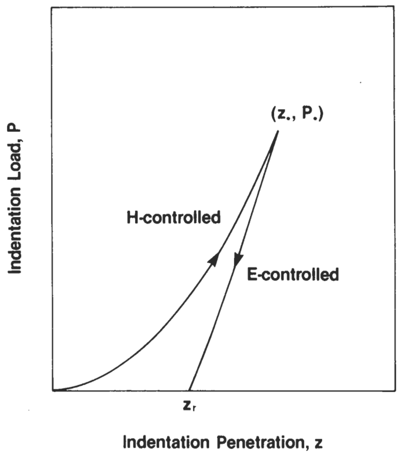

Fig. 2. Schematic of load-displacement response for sharp indenters.

indentation cycle, Fig. 2. On the grounds of geometrical similitude it can be argued that the contact pressure ($\propto P/a^2$) should remain constant through each half-cycle, in which case the load would be expected to relate to penetration ($z \propto a$) in some quadratic fashion. During the loading half-cycle the deformation has both elastic and inelastic components, and the contact pressure is (by definition) determined by the hardness; during unloading (and any subsequent reloading) the deformation is entirely elastic, and the contact pressure is determined by Young's modulus. The appropriate functional relations for the two half-cycles are[18]

$$P \propto Hz^2 \qquad \text{(load)} \tag{1a}$$

$$P \propto E\left(z^2 - z_r^2\right) \qquad \text{(unload)} \tag{1b}$$

where due allowance is made in the latter for the existence of a residual impression depth z_r at the completion of indentation. The requirement that these two relations be compatible at the maximum penetration z_* then leads to the result

$$\left(z_r/z_*\right)^2 = 1 - \eta H/E \tag{2}$$

where $\eta \simeq 6$ for Vickers indenters.[18] For glass, using the value of H/E quoted above, we compute $z_r/z_* \simeq 0.6$, indicating that the springback effect is indeed substantial.

An advantage of the formulation in Eq. (1) is that $P(z)$ may be readily integrated to determine the energy expenditure during indentation. The calculation of thermal effects at high rates of loading is one application of

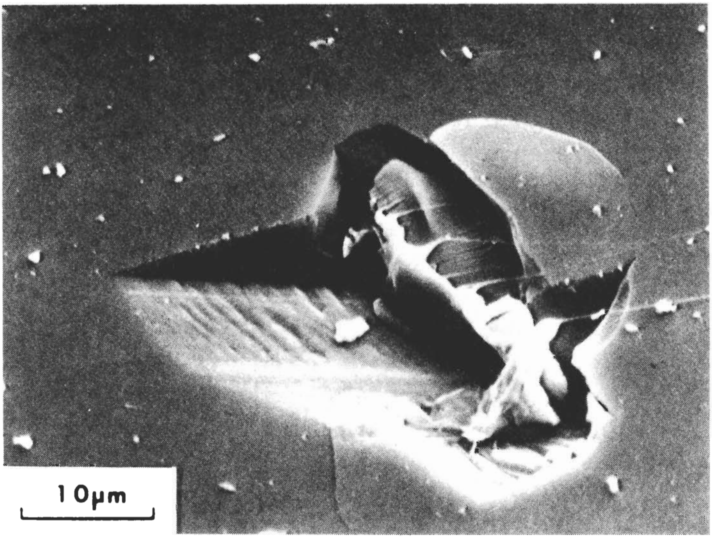

Fig. 3. Scanning electron micrograph of damage on soda-lime glass surface, produced by impacting silicon carbide particle (150 mesh size) at (glancing) incident velocity 50m s^{-1}. Note "molten" zone at surface. Courtesy B. J. Hockey.

such energy calculations. Under impact loading, for instance, adiabatic heating conditions are expected to prevail in the deformation zone. The question arises; is this heating sufficient to influence the local mechanical response of the material? Most quantitative evaluations of the effect in the past have concluded that it is not. However, these same earlier evaluations were all oversimplistic in one important respect; they assumed implicity that the thermal energy as generated *uniformly* within the final deformation volume. In reality, most of the irreversible work of indentation must be dissipated in the surface regions close to the initial contact point, for the more remote regions near the outer zone boundary undergo plastic work in only the later stages of loading. When proper account is taken of this factor in the energy equations we obtain the following expression for temperature rise in terms of radial distance r from the contact point:[19]

$$\Delta T = 3\theta \ln(b/r) \tag{3}$$

where θ, which involves such quantities as hardness, density and specific heat, is the equivalent temperature rise for uniform dissipation. An appropriate evaluation for soda-lime glass gives $\theta \simeq 133°C$, so it is clear that relatively high temperatures can be achieved at small r. Indeed, the level of heating can be sufficient to "melt" the glass at the contact surface, as is evident from the micrograph of a particle impact site in Fig. 3.

It will be noted that to this point our treatment has been purely phenomenological; we have not yet invoked a constitutive law for the deformation. Indeed, we have not even identified the *form* of the deformation; and with good reason, for herein lies one of the great areas of controversy in glass science.[20] Is the underlying process essentially shear or pressure activated? If the former, which basic laws, those of plasticity or those of viscosity, more properly describe the response?; it is certainly a fact that the hardness can exhibit strong rate effects.[21] Densification is known to be important in silica-rich glasses,[20] where the absence of network modifiers makes flow more difficult, but to what extent does this mode govern the indentation behavior in normal glasses? It is clear that questions like these will need to be answered if we are to understand how the fracture patterns generate. These are issues to which we shall return in the final sections of this paper.

There is nevertheless one general conclusion of far-reaching significance that we may draw from our simple approach: a state of residual stress must exist both within and without the deformation zone. The mere existence of the hardness impression is sufficient ground for asserting that this must be so. It is true that if viscous flow were to operate these stresses would relax with time (a process enhanced, of course, at elevated temperatures, as in annealing); or, alternatively, if densification were to occur, that the compacted zone might be more easily accommodated within the elastic matrix, thereby producing a less intense field in the first place.[10] However, such details concern only the question of degree, and we shall find the residual field emerging as a primary driving force in the ensuring fracture.

B. FRACTURE PATTERN

a. INITIATION The cracks shown in Fig. 1 appear only above some threshold in the loading. Below this threshold the indentation pattern has all the outward appearances of a plain hardness impression; above the threshold the cracks are well developed. This identifies an initiation stage in the fracture process. Direct observation of the contact site during actual testing reveals a strong tendency for the pop-in to occur on unloading, consistent with a dominant residual stress influence. The fact that this same pop-in behavior is observed on even pristine glass surfaces[22] indicates that the source of the initiation must be created by the deformation process itself.

Regardless of the source of the crack nucleus, an instability condition for the threshold can be determined using equilibrium fracture mechanics.[23] It is assumed that an incipient microcrack will become critical when its driving force exceeds the intrinsic toughness, K_c. Since the intensity of the indentation field is load-invariant (as reflected in a constant hardness), this critical driving force has to be achieved by increasing the spatial extent (e.g.

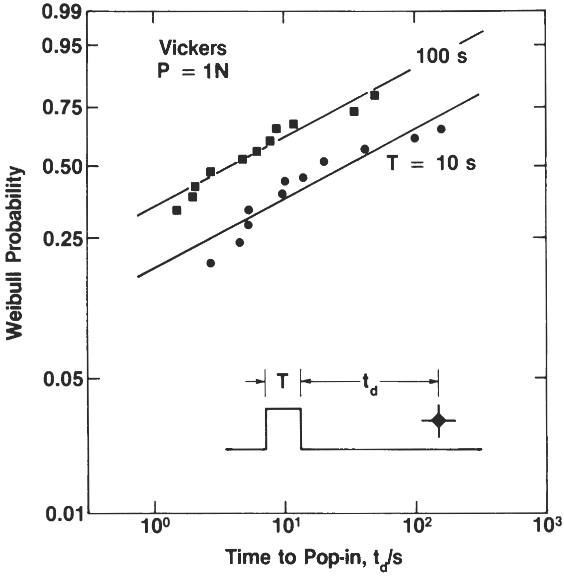

Fig. 4. Weibull probability diagrams showing delay times for radial crack pop-in for soda-lime glass in air. Data for fixed indentation load, at two hold times. Data courtesy C. J. Fairbanks.

as measured by the deformation zone size) of the field. On the basis of this argument the threshold load is found to be[23]

$$P_c = A(E/H)(K_c/H)^3K_c \tag{4}$$

where A is a numerical constant (slightly different for each crack system in Fig. 1[24]). For silicate glasses $P_c \approx 10N$ under inert test conditions, a load easily attained in spurious contacts with sharp particles in everyday handling. It is easy to see from Eq. (4) why glasses (and indeed most ceramics) are so brittle in comparison to metals; the latter class of materials is tougher by at least an order of magnitude and can consequently withstand relatively enormous stress concentrations without crack generation.

In practice, it is difficult to achieve the equilibrium fracture conditions assumed in the derivation of Eq. (1). Chemical species in the environment, notably water, can lead to a substantial reduction in the threshold loading; moreover, this reduction increases as the time of contact is prolonged, thereby introducing a kinetic element into the initiation.[25] The most dramatic illustration of the rate effect is seen at load levels just *below* those needed to produce cracks *during* contact, whence pop-in occurs *after* contact, once more demonstrating the effectiveness of the residual field. This delayed fracture event is subject to considerable variability in its kinetics, as is evident from the Weibull plots in Fig. 4 for radial cracking in soda-lime glass. Figure 5

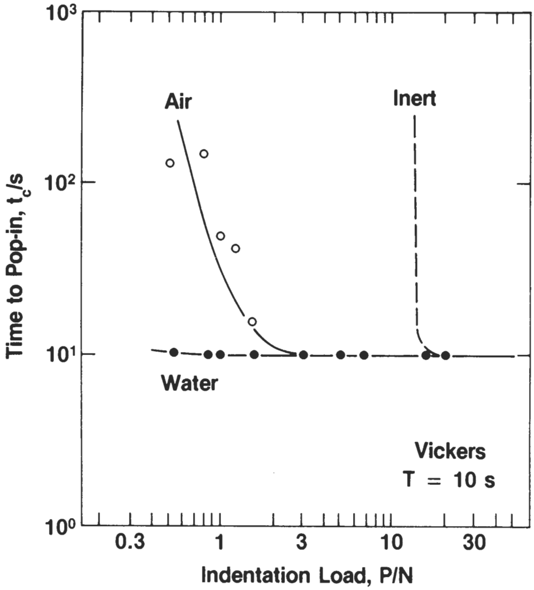

Fig. 5. Median times for radial crack pop-in as function of indentation load for soda-lime glass in three environments. Time axis plotted as $t_c = T + t_d$ (see Fig. 4) to allow condition $t_d = 0$ to be represented on diagram. Data for fixed indentation hold time. Data courtesy C. J. Fairbanks.

shows how the median delay time varies with indentation load for the same glass, at a given hold time, in both air and water. The chemical influence on the development of a critical nucleus is clearly strong. Whether this influence is manifested via the deformation processes which nucleate the microcrack precursors to pop-in, or via subsequent subcritical growth of the microcracks themselves, is a question which is only now being addressed.[25] We shall raise this question again later when we investigate the mechanistic aspects of the indentation of glass.

More recent work at N.B.S has confirmed that the delayed pop-in phenomenon is by no means specific to the one glass represented in Figs. 3 and 4.[26] Even fused silica, with its predominatly compactive mode of deformation (Sect. IIA), exhibits the effect, albeit not so strongly. Another finding is that the surface history of the glass can be an important factor in the kinetics. It would seem that there is much yet to be learned from such studies.

b. PROPAGATION Once the initiation barrier has been overcome, the cracks propagate into their ultimate well-developed configurations. One feature

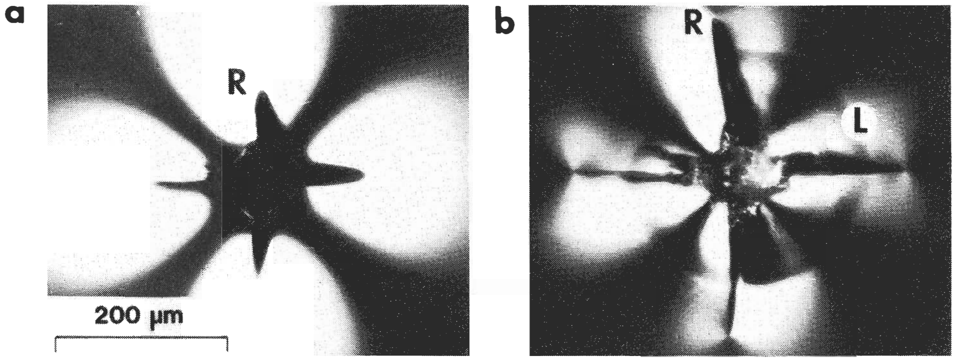

Fig. 6. Optical micrographs of Vickers indentation in soda-lime glass viewed from below contact (a) at full load (P = 40N) and (b) at complete unload. Crossed polars. Radial and lateral cracks denoted by R and L, respectively. After Ref. 10.

common to all such final configurations, regardless of whether the crack is of the radial or lateral (or even cone) type, is an effective penny-like geometry; that is to say, the unloaded cracks have circular fronts. However, this apparently simple geometry belies the complex growth history that takes place enroute to these configurations, particularly in the case of the radial system. The elastic component of the stress field which drives the radial cracks is not symmetrical about the contact center; whereas it is tensile *below* the surface, it is compressive *at* the surface.[11,12] Consequently, if initiation can be induced during the first half-cycle of the indentation, as it can at loads well above the threshold, propagation takes place in the downward direction. Radial extension along the surface occurs only during the unload half-cycle, as the elastic "constraint" is progressively removed. The inelastic component of the field is, by contrast, more or less uniformly tensile in all directions, thereby accounting for the ultimate radial symmetry of the crack front. These evolutionary features are apparent in the micrographs of Fig. 6, showing subsurface views of a Vickers indentation at full load and complete unload in soda-lime glass. The lateral crack system (seen faintly in Fig. 6) is subject to an even stronger elastic constraint, and hence develops even later in the cycle.[13]

The fact that the crack systems continue their development up to the end point of the contact process indicates that the immediate post-indentation configurations correspond to equilibrium states. Under these conditions, and assuming that the residual driving force is concentrated at the center of contact (far-field approximation, c >> b, Fig. 1), we find

$$P/c_r^{3/2} = B(H/E)^{1/2}K_c \qquad (P > P_c) \tag{5}$$

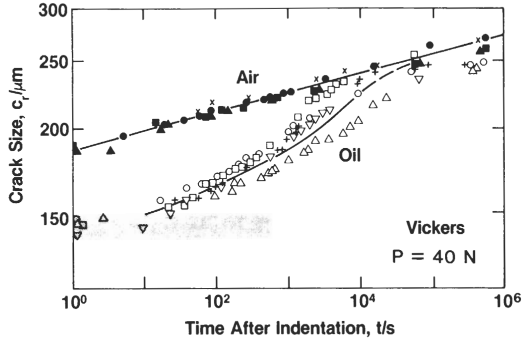

Fig. 7. Post-indentation radial crack growth as a function of time in soda-lime glass. Data for several cracks at given indentation load in two environments. Shaded band indicates immediate post-indentation crack size. After Ref. 28.

for the radial geometry, where B is a numerical constant. Typically, $P/c_r^{3/2} \approx 30$ MPa.m$^{1/2}$ for silicate glasses, so cracks at the inert-environment threshold load (previous subsection) will pop-in to $c_r \approx 30\ \mu$m. This value is characteristic of naturally occurring flaws at the top end of the size distribution. An analogous, albeit somewhat more complex, formulation is available for the lateral crack geometry.[13] In inverted form, Eq. (5) serves as a formula for evaluating toughness parameters directly from the radial crack traces.[27,28]

As with the initiation stage, the assumption of equilibrium conditions in the crack propagation may not always be closely realized. Once again, moisture in the environment can have a significant influence, especially in the post-indentation configurations. The newly formed cracks remain under the influence of the residual stress field (witness the intense stress birefringence in Fig. 6b), and are therefore subject to subcritical growth.[28,29] This post-indentation growth is rapid at first, but slows down as the fronts propagate further outward from the central driving force. The extent of the growth can be substantial, as seen in the data for soda-lime glass presented in Fig. 7,[28] although again it is less pronounced in those "anomalous" glasses which deform by densification rather than by flow.

III. MECHANISMS OF DEFORMATION AND FRACTURE

Thus far we have confined ourselves to descriptions of the indentation process in the language of engineering mechanics. These descriptions are useful for design purposes, but they provide little physical insight into the nature of the underlying deformation and fracture processes. It will be recalled that the scale of events we are concerned with here is of the order of micrometers, so to gain such insight we must expect to have to resort to high resolution microscopy techniques.

A. DEFORMATION

Over the past decade or so some interesting revelations have been made, primarily by Hagan and co-workers, on the mechanisms of indentation deformation in glass.[10,25,30–33] The crucial finding is that, in glasses which deform by "normal" volume-conserving flow, the primary mechanism is highly inhomogeneous within the contact zone. Essentially, the flow is localized on well-defined, shear-activated fault interfaces, much as classical dislocation slip processes in crystalline materials. These fault interfaces are clearly discernible by the traces they leave on section planes through the deformation zone; Fig. 8, an S.E.M. micrograph of an indentation in an arsenic trisulphide glass, is an illustrative example. The development of the fault patterns may be pictured in terms of an intermittent "punching" mode, whereby catastrophic slip occurs at intervals along shear stress trajectories to accommodate the intense strains (close to the theoretical limit sustainable by the molecular structure) imposed by the penetrating indenter.[25]

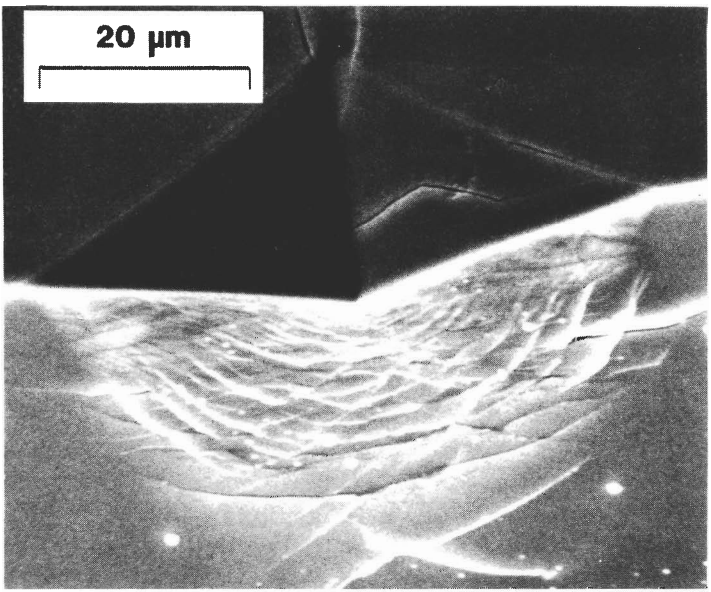

Fig. 8. Scanning electron micrograph of Vickers indentation (P = 2N) in arsenic trisulfide glass. Surface plus section view (latter obtained by indenting across a pre-existing crack and breaking specimen). Courtesy T. P. Dabbs.

From a detailed analysis of the observed shear fault characteristics the following pertinent conclusions may be drawn:[25]

(i) The newly formed fault walls tend to recontact and heal under the action of highly compressive stresses within the contact zone. However, the penetration of water into the interfaces can cause decohesion by hydrolytic reactions with strained silanol bonds, thereby increasing the local compliance of the glass structure. This decohesion process would account for the rate-dependent reductions in hardness values for glasses tested in moist environments.[17]

(ii) Faults generated in different quadrants of the Vickers contact area can intersect either at or below the glass free surface, thereby creating (sessile) centers of high stress concentration.[33] These centers constitute favorable sites for crack nucleation and growth. The spatial distribution of such centers within the contact zone is expected to show a certain variability, consistent with the intermittent nature of the shear events.

Analogous microscopic studies of indentations in anomalous glasses, on the other hand, reveal comparatively structureless deformation zones.[10,32] There is an implication here of a less disruptive mechanism. Ernsberger[20] and others account for this different behavior in terms of a change from a predominantly *reconstructive* mode (as must be characteristic of the flow mechanism discussed above) to a *displacive* mode (characteristic of the simplest kind of phase transformation) of molecular rearrangement.

B. FRACTURE

a. INITIATION The preceding description leads us to the notion of crack initiation as a two-step process, precursor shear faulting followed by tensile mirocrack development. Fine details of this process, especially in connection with the micromechanics of crack nucleation at the fault intersections, remain obscure, but we are nevertheless able to offer explanations for some of the broader features of the threshold phenomenon.[25]

Thus, according to our model we should expect that, under equilibrium conditions, the first step in the initiation should be complete at maximum penetration of the indenter, for it is at this point in the cycle that the shear displacements reach their maximum. Under non-equilibrium conditions it is possible that this precursor phase may continue in its development during the ensuing unload stage of the contact; water-induced decohesion can lead to a build up of stress intensification at the obstructed faults, which, depending on the unload rate, outweighs the countervailing effect of the reduced shear displacements. The second step in the initiation can take place only when the microcrack nuclei experience tensile driving forces. We have already indicated that the elastic component of the field tends to preclude the development of tensile stresses until toward the end of contact, at least in the near-surface regions. (We recall from Fig. 6 that, at loads well above the threshold level, it is possible to develop cracks in the subsurface regions earlier in the cycle.) Once tensile stresses are felt, the nuclei can evolve further by moisture-assisted subcritical crack growth. Qualitatively, therefore, our proposed two-step picture would appear to have the capacity to account, in a self-consistent way, for both equilibrium and kinetic characteristics of the pop-in event.

A deeper understanding of the initiation problem than the somewhat superficial picture we have just painted awaits the development of *quantitative* modelling. This is particularly true of the delayed radial fracture behavior described earlier (Figs. 4,5). Which of the two steps above exerts a rate-controlling influence in the instability micromechanics? How might the elements of variability in the locations of nucleation centers within the deformation field be accommodated into an appropriate distribution function to account for the large scatter in the observed delay times? How far might we expect to be able to extend the two-step concept to anomalous glasses, where the delay-time phenomenology is apparently similar but the primary mode of deformation is demonstrably different? Until we have answers to such questions the basis for designing with glasses in the region of ultra-high strengths (i.e. near the theoretical limit) will remain entrenched in empirical data-fitting procedures.

b. PROPAGATION We have indicated that the deformation processes in indented glass can play a vital role in crack initiation, by creating the incipient nuclei. In this sense the deformation is clearly an integral part of the very mechanism of crack formation. It has been suggested that a similar intimate inter-relationship may be evident in the propagation stage of fracture as well.[16,21] There is no doubt in our case that the deformation does have a strong influence on the propagation, via the residual driving force. However, this force operates at the mouth of the crack and not at the tip, and hence affects the mechanics rather than the mechanisms of growth. Indeed, newly propagated cracks in glass can be made to heal spontaneously upon removal of the driving forces,[34] rather like a reversible "zipper," thus providing compelling (if circumstantial) evidence for the classical picture of fracture by sequential bond rupture. Simulations of crack-tip structures in brittle materials by computer modelling[35] support this picture, indicating that cracks can indeed grow without plastic instabilities, and, moreover, that such cracks are atomically sharp.

Of course, truly definitive evidence on the mechanisms of material separation at crack tips can only be obtained by direct observations of the tips themselves. Unfortunately, the small, near-atomic scale predicted for the critical tip processes makes such observations extremely impractical. The one general technique which has the necessary resolution capability, transmission electron microscopy, cannot be used in its powerful diffraction contrast mode in glasses. However, extensive TEM studies on tip regions of indentation cracks in crystalline solids by Hockey[36–38] would seem to bear on the general nature of brittle fracture. Thus in the micrograph of Fig. 9 showing a radial crack segment in aluminum oxide there is evidence of the same kind of reversibility, from the healing dislocation network at the original crack interface,[37] as alluded to above for glass. In this case, there is no indication that slip dislocations have emitted from the tip region at any stage of its development. Even in relatively soft ceramics with strong dislocation activity, such as magnesium oxide, the tip regions of freshly propagated cracks appear, without exception, to preserve their essentially sharp configuration.[38]

This issue of crack-tip geometry is worthy of consideration, for it calls into question the fundamental laws of fracture. There are some who contend that truly brittle cracks do not have the quality of atomic sharpness, but rather are rounded at their tips.[39] One of the disadvantages of any theory based on the latter notion is that it becomes necessary to incorporate a curvature parameter into the criteria for fracture. That is to say, it no longer remains possible to express crack laws uniquely in terms of stress intensity factors [e.g., $K = K_c$, under equilibrium conditions; or, $v = v(K)$, under kinetic conditions, where v is velocity], in which case much of the power and convenience of linear

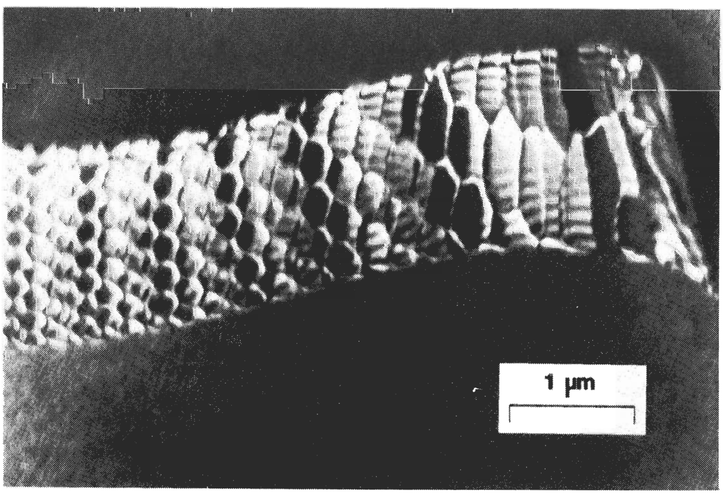

Fig. 9. Transmission electron micrograph of radial crack segment in aluminum oxide showing healing dislocation network at interface. Courtesy B. J. Hockey.

elastic fracture mechanics is lost. At the same time, our implied correspondence between the concepts of ideally sharp crack and stress intensity factor has its own limitations, for the K-field has a singularity at its origin; we need to remember that the critical crack-tip processes are tied up with the characteristics of interatomic forces, and are therefore essentially nonlinear and discrete in their nature. Furthermore, it should be made clear that this same implied correspondence does not preclude the possibility that truly brittle cracks may, at some stage in their lifetime, become blunted; or, that some strength-controlling flaws may actually begin their evolution to failure with rounded contours. In such instances the "crack" takes on the quality of a notch, and the distinctions between initiation and propagation become blurred.[40] These topics will be taken up in greater detail by others at this meeting.

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DISCUSSION

Q: W. B. Hillig. Can you elaborate on the need for the critical load for fracture in view of the geometrical similitude in the Vickers indentation arrangement?

A: B. R. Lawn. Both fracture and deformation may satisfy the principles of geometrical similitude in indentation, but in different ways. Whereas fracture is essentially an area-dependent process, deformation is volume dependent. Thus the energy expenditure involves the square of some characteristic linear dimension in the first case, and the cube in the second. Hence there is an inherent trend from surface-dominated behavior at small indentations to bulk-dominated behavior at large indentations, a phenomenon common in science. Our threshold load may be loosely interpreted in this way.

Q: G. H. Frischat. In some of your formulas you used the quantity H/E and you estimated it to be about 0.1 for "normal" glasses. However, H/E is not constant at all if you use the conventional hardness H. It may vary by almost a factor of 10 as a function of load (J. Mater. Sci. *17*, 329, 1982).

A: B. R. Lawn. It is true that our formulas are based on the assumption that H/E remains a constant, and that this constancy is not always satisfied. However, the derivations referred to above generally occur at low indentation loads, below about 1 N. Most (not all) indentation testing is carried out at higher loads than that.

Q: M. Tomozawa. On the crack propagation, you excluded the possibility of plastic deformation. But don't you think that plastic deformation at the crack tip is possible, or probable, when water (or moisture) interacts with the crack tip?

A: B. R. Lawn. As mentioned in the paper, there is compelling evidence, from direct observations of crack tips in several materials, using transmission electron microscopy, that brittle cracks propagate by a bond rupture mechanism. The materials studied, although crystalline in structure, exhibit the same fracture phenomena as glass. Thus I believe that plastic deformation is not the essential crack-tip separation process responsible for fracture, at least under normal propagating conditions. This is not to say that plastic flow could never occur at crack tips. A mark of such a process would be the occurrence of irreversible effects in the propagation characteristics, as for example observed by Michalske in his "crack blunting" experiments (this volume).

Q: W. B. Hillig. It seems to me that all three of the crack tip mechanisms can operate depending on material and/or environment. Plastic deformation can occur in some amorphous solids such as amorphous metals and perhaps chalcogenide glasses. Rounding implies either material flow or material removal, such as due to a reactive environment. In the absence of a reactive environment that removes, or transforms material to a non-load bearing region, then the crack tip will be atomically sharp.

Q: T. A. Michalske. Do you think that it is productive to draw distinctions between "sharp" and "plastically deformed" crack tips? In the atomic region near the crack tip, very high strain levels may result in deformations which do not conform to macroscopic properties of pure elastic or plastic behavior.

A: B. R. Lawn. I think it is productive to make such distinctions in the sense that we may understand the basic nature of crack propagation, and thence make predictions about how cracks will behave in different conditions. Thus, if we identify bond rupture as the essential crack growth process we are lead to seek models of fracture in terms of the properties of the atomic bond. This has led to highly successful predictions concerning the interactions of cracks in certain materials with specific chemical environments (S. W. Freiman, this volume). If we were to identify plasticity as the essential growth process we should have to look to bulk deformation processes to provide all our answers. The distinction is not one of semantics.

Q: J. J. Mecholsky. Is it possible that the "shear faults" are a result of cracking beneath the indenter at the region between the high compressive stress and the tensile stress. The cracks that originate at this boundary (due to the tensile stress) change the stress field such that now a second set of cracks can form parallel to the existing cracks and closer to the indentation impression. This new set of cracks changes the stress field such that new cracks develop parallel to the last, etc. Is there any evidence to dispute this development of the so called "shear faults"?

A: B. R. Lawn. As yet, the way the shear faults form during indentation remains obscure. All we can say at present is that *after* the event the faults lie on surfaces which closely delineate shear stress trajectories in the contact field. It is possible that the explanation offered above is correct, but this remains to be established.

Q: H. Schaeffer. I wonder which glasses fall into which category? Is it as simple to state that "anomalous" glasses (showing predominantly densification effects) are the alkali-free more refractory glasses, whereas the "normal glasses" (predominantly plastic effects) are the ones with little network cohesiveness (as an extreme autoclaved glass)? Would temperature increase transform "anomalous" to "normal" glasses?

A: B. R. Lawn. No, I do not think the distinction between network modifier and network former allows us to make a complete distinction between normal and anomalous glasses. The answer to the more specific parts of the above question would require more knowledge of structural properties than possessed by this author. Nevertheless, the distinction does appear to be very real and worthwhile.

Q: F. Geotti-Bianchini. (i) How deep under the original surfaces do lateral cracks nucleate?

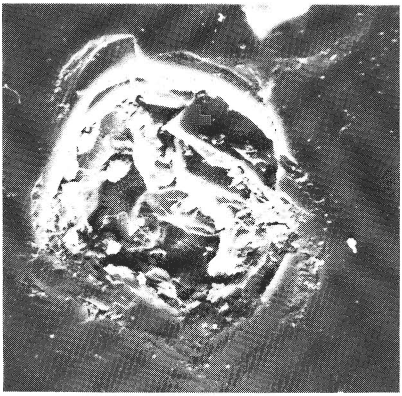
(ii) How does this reflect on the indentation behavior of glasses with surface coatings (a few tens of A° to 5000 A°)?

A: B. R. Lawn. (i) The lateral cracks appear to nucleate in subsurface regions close to the base of the deformation zone, i.e. typically several tens of micrometers.

(ii) I would be surprised if thin films were to have any significant effect on lateral crack nucleation, although they might on radial crack nucleation, which occurs closer to the indented surface. The film might play a role by impeding access of the environment to the crack nuclei, or by modifying frictional stresses of the indenter-specimen interface.

Q: P. K. Gupta. (i) It appears from some of your micrographs that crack initiation occurs within the shear faults. Do you think it is more reasonable that initiation occurs in faults and then the rest of the process of crack evolution is essentially propagation?

(ii) The following SEM micrograph of a Vickers indentation on a fused silica (Dynasil) shows evidence of shear faults and even radial cracks. The top figure shows these more clearly. I thought you mentioned that silica



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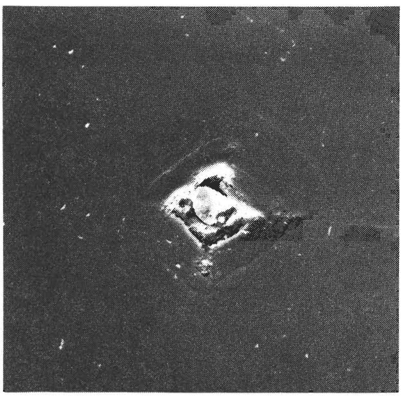

600 x

does not show shear faults and instead densifies. Do you agree with me that the attached micrograph shows shear faults and radial cracks? If you do, then what additional comments do you have?

A: B. R. Lawn. (i) Although the microscopic evidence associating the cracks with shear faults is compelling, detailed mechanisms of the initiation process remain obscure. It is indeed conceivable that initiation does occur within the interfaces of the faults themselves, but that it is a possibility which awaits experimental confirmation.

(ii) Silica glass does tend to densify rather than deform by classical plastic flow, but this does not mean that shear faults are totally absent in this glass type. I do agree that the S.E.M. micrograph shows surface markings, but I do not think these features follow shear trajectories below the surface. Our observations (and those of Hagan) do not reveal the characteristic shear line markings in section views apparent in soda-lime glass. Nevertheless, radial cracks do occur at indentations in silica; moreover, these cracks are driven by the same residual stress fields (albeit at reduced intensity). We take up these points in greater depth in later papers in this volume.

C: J. T. Hagan. (written contribution) The shallow surface cracks within and at the corners of the indentations in fused silica in P. K. Gupta's micrographs are definitely not traces of shear faults. The subsurface deformed zone in fused silica does not show any shear flow or fault lines and this can be confirmed by sectioning the indentation.[1,2] These surface cracks are similar in origin to the surface ring cracks associated with Hertzian elastic contacts except that they reflect the symmetry of the Vickers indentation; they arise from the high elastic contact stresses associated with wide angle wedge indentations in highly elastic solids.[3]

The radial cracks from the corners of the indentations in fused silica are formed by unloading residual stress and have nothing to do with shear faults. The initiation at the corners of the indentation is an artifact from the stress concentrations at the corner of the indenter/specimen contact.

Crack initiation or void formation can occur within the shear faults which only occur in normal glasses such as soda-lime glass.[4] In such glasses, the shallow surface cracks within the indentation may be traces of 'ring'-type cracks which link up with subsurface flow lines.

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C: R. H. Doremus. I agree with your caution in calling the shear faults either cracks or slip lines. The shear faults have unique characteristics that make them a separate mode of deformation. They are not cracks because they hold together to some extent, and they are not like the dislocation motion that causes slip in crystals. Dislocations move at low stresses and can leave no trace behind; the shear faults only occur at high stresses and leave a narrow (a few atomic distances) distorted zone.